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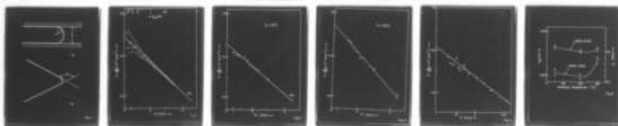
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EVIDENCE FOR IDEAL ELASTIC PLASTIC DEFORMATION
IN Fe-Ni BASED METALLIC GLASSES.

by

⑩ D. G. Ast ~~and~~ D. J. Krenitsky

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⑫ 18p.

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It is shown that metallic glasses are ideal elastic plastic solids up to strains of about 7.5% . Yield stress variations of 20% from batch to batch are most likely caused by variation in the processing.		

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ABSTRACT

It is shown that a Ni-Fe based metallic glass ($\text{Fe}_{40}\text{Ni}_{40}\text{P}_{14}\text{B}_6$) behaves in bending as an ideal elastic plastic solid. The result implies that the slip distribution in shear bands remains unchanged when the applied stress is removed.

Batch to batch variations in the yield stress of 20% are most likely caused by differences in the preparation conditions.

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1. Introduction

The stress σ vs strain ϵ relation of metallic glasses is not easily investigated because deformation in tension is confined to a very small fraction of the sample (typically one shear band) resulting in a macroscopic brittle behavior. Compression tests in Pd-Si based glasses, where relatively modest quenching requirements allow the production of bulk specimens, indicate that these materials behave essentially as ideal elastic plastic solids.¹ The stress strain curve of Fe-Ni based metallic glasses, which can only be prepared in the form of thin ribbons has not been investigated to date. Bending tests are seldom used to derive stress-strain data but have the important advantages that the deformation is stable, even when $\partial\sigma/\partial\epsilon \leq 0$, and that no bulk specimens are required. In this paper, we will analyze bending tests on Fe-Ni based metallic glasses and demonstrate that the material behaves within narrow experimental limits as an ideal elastic plastic solid.

2. Materials and Methods

The metallic glass was purchased in 1976 from Allied Chemical in the form of a continuous thin ribbon of 51 μ m thickness and 1.60 mm width. The trade name is 2826, and the composition given by the manufacturer as $\text{Fe}_{40}\text{Ni}_{40}\text{P}_{14}\text{B}_6$ agrees within experimental error of about 1% with an analysis carried out in our laboratory. A second roll of 2826 was purchased in 1978. This material was slightly thicker (53 μ m) and somewhat narrower (1.47 mm). Although of identical composition, the mechanical properties of this alloy are, as we will show later, considerably different from the 1976 alloy. The material was cut into approximately 6 cm long sections which were slowly compressed between parallel platens

as shown in Figure 1a. At platen spacings (D) smaller than about 3 mm, this leads to a permanent deformation characterized by a kink angle α (see Figure 1b). The relation between D and α was measured for as received material and for material annealed at 100°, 150° and 200°C.

To compress the material, we used a pair of machinists calipers which could be read to ± 0.01 mm. α was measured with a protractor to about $\pm 1^\circ$.

3. Analysis and Results

The data were analyzed with a model which was based on the following assumptions:

- a. Between the platens, the ribbon is bent into a semi-circle of radius $R = (D-d)/2$, where d is the thickness of the ribbon. (See Figure 1a).
- b. The stress strain curve of the material consists of two linear sections which represent the elastic and plastic part respectively.

The initial value of the flow stress in the plastic region, σ_y , and the slope, m, of σ vs ϵ_p are adjustable parameters to be determined from experiment. Negative, zero and positive values of m correspond to work-softening (linear in ϵ_p), ideal plastic, and work-hardening (linear in ϵ_p).

A semicircle is only a first order approximation of the true shape of the ribbon between the plates. For example, if one wishes to calculate the forces on the platens during compression one has to consider more accurate solutions (see Reference 2). However, in the case at hand where one is interested in the permanent deformation after compression, the semi-circular approximation is sufficient, since positive (more curvature) and negative (less curvature) deviations between the real and assumed shape cancel to first order.

The calculation of α vs D is straightforward, yielding

$$\left[\frac{1-(\alpha/180)}{D'} \right] = \frac{-\sigma_y^3}{2 \cdot E^3 \cdot d^3} \left(1 + \frac{m \cdot \sigma_y}{E} \right) \cdot D'^2 - \frac{m \cdot \sigma_y}{E} \cdot \frac{1}{D'} + \frac{3 \cdot \sigma_y}{2 \cdot E \cdot d} \left(1 + \frac{m \cdot \sigma_y}{E} \right) \quad (1)$$

where $D' = D-d$, α the kink angle in degrees, and E is the elastic modulus of the material. In the absence of work-hardening or -softening, Equation (1) reduces to:

$$\left[\frac{1-(\alpha/180)}{D'} \right] = - \frac{1}{2} \left(\frac{\sigma_y}{E \cdot d} \right)^3 \cdot D'^2 + \frac{3}{2} \left(\frac{\sigma_y}{E \cdot d} \right) \quad (2)$$

i.e., for an ideal elastic plastic solid, $[1-(\alpha/180)]/D'$ is a linear function in D'^2 and both the slope and the intercept depend on $(\sigma_y/E \cdot d)$.

The experimental values of $\alpha(D')$, as measured on 1976 material, are plotted as $[1-(\alpha/180)]/D'$ vs D'^2 in Figures 2, 3 and 4 for untreated and 100 and 150°C annealed material. 200°C annealed material was too brittle to be measured over a sufficiently wide range. The top horizontal scale in all figures indicates the macroscopic plastic surface strain, $\epsilon_{p,s}$, at a given D' .

Inspection of these figures shows that Equation (2) seems to be obeyed within experimental error. A least square fit over the range $0.0048 < \epsilon_{p,s} < 0.05$ yielded a correlation coefficient, r^2 , of 0.975.

For a more stringent test of the applicability of the model we calculate σ_y both from the slope and from the intercept with the vertical axis using a value³ of 144.8 GPa ($21 \cdot 10^6$ psi) for E . If the theory underlying Equation (2) is applicable, then the values obtained must be identical within experimental error. The values for σ_y of as received 1976 material obtained from the least squares fit above were 2.170 GPa (314.8 ksi) from the intercept and 2.244 GPa (325.5 ksi) from the slope. These values differ by about 3.3% which is within the scatter expected from r^2 . The average value of σ_y is 2.207 GPa (320.15 ksi).

The experiments were repeated with the 1978 material (see Figure 5). For the as received material the value of σ_y derived from the intercept was 2.606 GPa (378 ksi) and from the slope 2.647 GPa (384 ksi). The two values differ by 1.6%. The average value is 2.63 GPa (381 ksi).

The difference in σ_y of the two materials is outside the experimental error which is mostly determined by fluctuations in d and estimated to be $\lesssim 10\%$.

To assess the sensitivity of the above analysis to deviation from the ideal elastic plastic behavior, we calculate via Equation (1) the case of $m = 2.5$; i.e., the case where the flow stress changes by 2.5% for a 1% increase in plastic strain. The result of this calculation is shown in Figure 2 as dashed lines. Note that we are considering comparatively small deviations from an ideal elastic plastic solid, since in crystalline solids, m is typically one order of magnitude larger. An inspection of Figure 2 indicates that in a metallic glass any deviations from the ideal elastic plastic behavior, if they occur at all, must be smaller than about $m \leq |1|$.

Table 1 summarizes the yield strength data on annealed and unannealed material. Figure 6 shows these σ_y 's as a function of annealing temperature.

The value of σ_y of untreated 1976 material of 2.207 GPa (320.15 ksi) agrees well with a previous determination for material from the same spool. In this earlier investigation, σ_y was determined from an analysis of the forces exerted on the platens during compression and found to be 2.206 GPa (320 ksi).² In these experiments the material was subjected at all times to an applied stress. In the present experiment, the material is first loaded, then unloaded. The good agreement between the two σ_y values indicates that no relaxation occurs in the shear bands when the applied stress is removed.

The slight changes in σ_y/E of about 5% upon annealing are likely within the accuracy of the experiment. However, a similar slight increase in σ_y was also observed in Reference 2. If real, the increase would be consistent with observations in both glassy polymers and crystalline Fe alloys, which indicates that treatments which decrease the fracture toughness, K_c , generally increase σ_y . In 2826, K_c falls rapidly with annealing at $T \gtrsim 130^\circ\text{C}$.⁴

Microscopically, the deformation is carried by shear bands. In bending tests, these shear bands are relatively closely spaced (6 μm would be a typical

value) and have surface offsets of about 0.1 μ m. Thus metallic glasses behave under bending not more inhomogeneously than, for example, neutron-irradiated fcc materials in tension. This averaging over the properties of individual shear bands is an important advantage of the bending test. Even though each shear band propagates as a kinetically controlled plastic instability,⁵ the overall plastic behavior is in equilibrium with the applied stress.

The fact that these shear bands do not recover part of the plastic deformation upon removal of the applied stress was confirmed by interferometric observations of the surface and measurements of the slip distribution in first stressed and then unstressed shear bands.⁶ From these observations one can estimate that the flow stress in sheared and non-sheared material must be identical to within $\approx 5\%$. As far as the flow stress is concerned, therefore, deformed and undeformed material are essentially identical. An interesting question is how this finding can be incorporated into models which assume that shear deformation leads to permanent (at RT) structural difference between undeformed and sheared material; i.e., differences in the amount of short range order⁷ or amount of free volume.⁸ Quantitative estimates of the effect of the postulated structural changes on σ_y would be valuable.

Models which consider shear band deformation by dislocation mechanisms⁹ predict basically similar mechanical properties of sheared and unsheared material since the passage of dislocations is not considered to introduce structural changes.

The process of shear band initiation, propagation and termination, however, is not well understood at present. Further work will be required before firm conclusions can be drawn.

The improvement in the flow stress of almost 20% between 1976 and 1978 indicates that preparation conditions have considerable influence on σ_y . The result

is not unexpected in view of the fact that the relaxation kinetics of Fe-based metallic glasses depend critically on the quenching history of the sample.^{10,11} It is possible that annealing at higher ($T \geq 200^\circ\text{C}$) temperatures (at which pronounced structural relaxations take place) will narrow the differences in the flow stress of variously prepared 2826. The trend in Figure 6 would support such a speculation but the anneal-introduced changes in σ_y/E are too small compared to experimental errors to draw any firm conclusions.

ACKNOWLEDGEMENTS

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Flow stress of $\text{Fe}_{40}\text{Ni}_{40}\text{P}_{14}\text{B}_6$ after various anneals

Annealing Temperature:		RT	100°C	150°C
1976 material	from slope:	2.244 GPa	2.186 GPa	2.347 GPa
	from intercept:	2.171 GPa	2.116 GPa	2.322 GPa
1978 material	from slope:	2.648 GPa	2.543 GPa	2.626 GPa
	from intercept:	2.606 GPa	2.581 GPa	2.573 GPa

Table 1.

FIGURE CAPTIONS

Figure 1. Approximate geometry of a ribbon compressed between parallel platens (a) and the resultant permanent deformation (b).

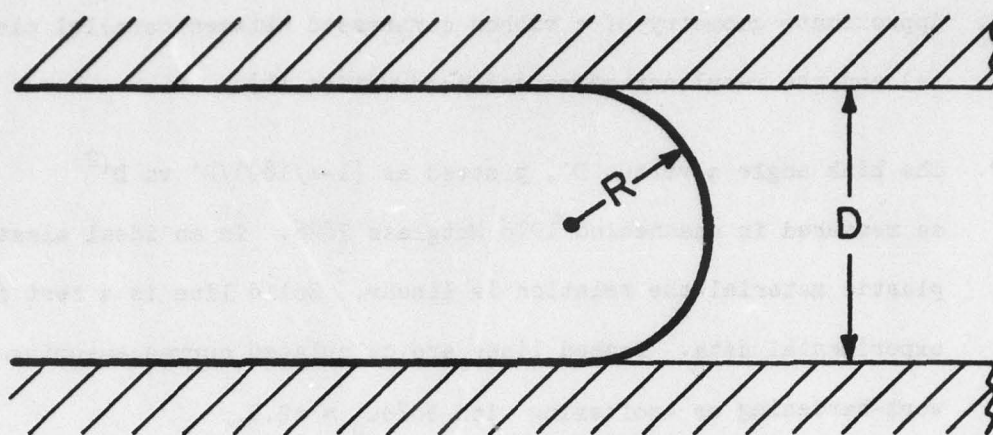
Figure 2. The kink angle α versus D' , plotted as $(1-\alpha/180)/D'$ vs D'^2 as measured in unannealed 1976 Metglass 2826. In an ideal elastic-plastic material the relation is linear. Solid line is a best fit to experimental data. Dashed lines are calculated curves assuming linear work-hardening or -softening with $\partial\sigma/\partial\epsilon_p = \pm 2.5$.

Figure 3. The kink angle α versus D' , plotted as $(1-\alpha/180)/D'$ vs. D'^2 for 1976 Metglass 2826 annealed for 30 min at 100°C.

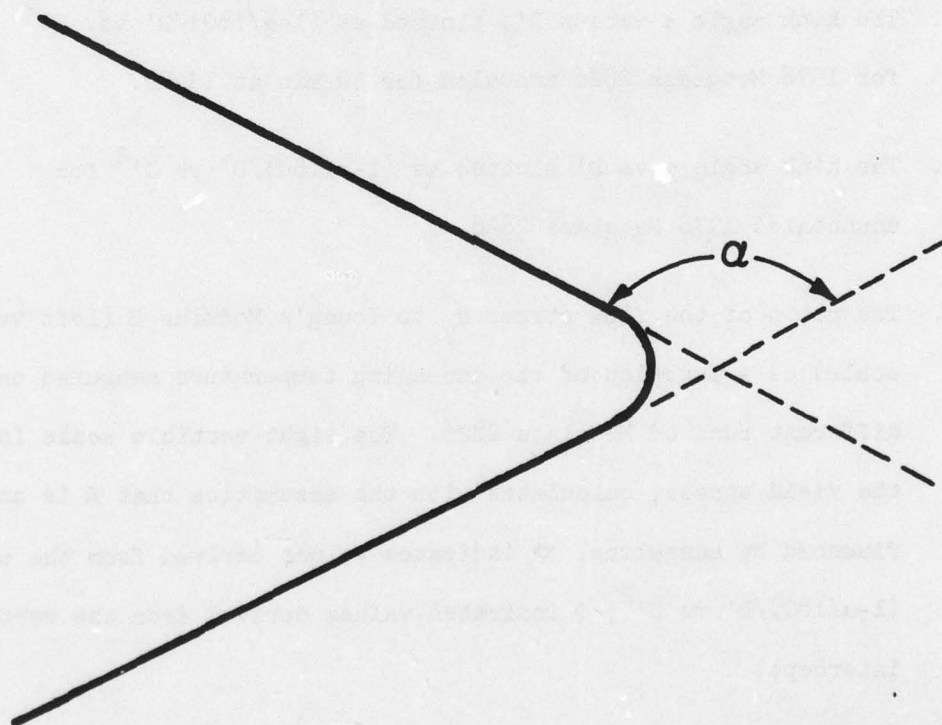
Figure 4. The kink angle α versus D' , plotted as $(1-\alpha/180)/D'$ vs. D'^2 for 1976 Metglass 2826 annealed for 30 min at 150°C.

Figure 5. The kink angle α vs D' plotted as $(1-\alpha/180)/D'$ vs D'^2 for unannealed 1978 Metglass 2826.

Figure 6. The ratio of the flow stress σ_y to Young's Modulus E (left vertical scale) as a function of the annealing temperature measured on two different runs of Metglass 2826. The right vertical scale indicates the yield stress, calculated with the assumption that E is not influenced by annealing. \diamond indicates values derived from the slope of $(1-\alpha/180)/D'$ vs D'^2 ; \circ indicates values derived from the vertical intercept.



a)



b)

Fig. 1

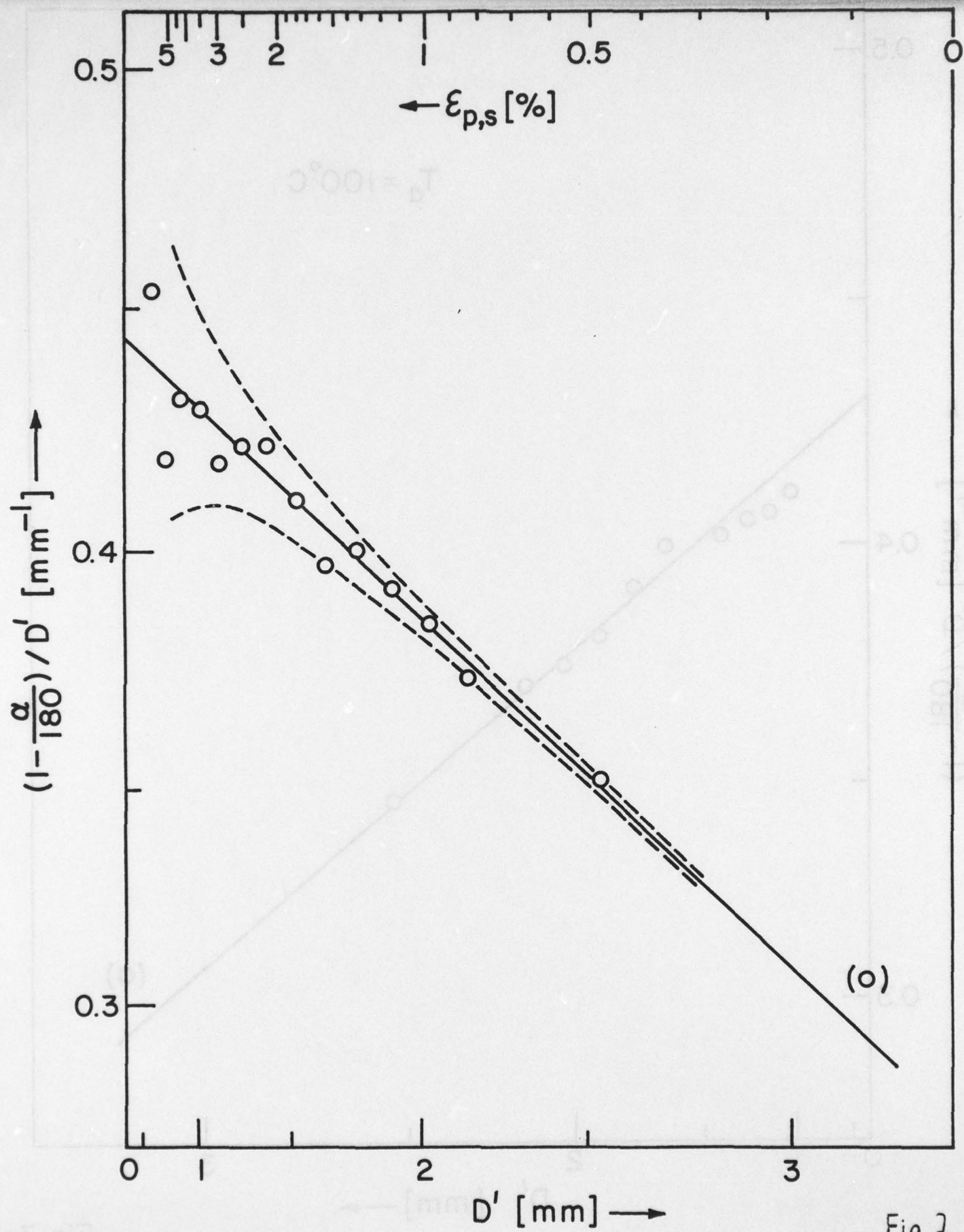


Fig.2

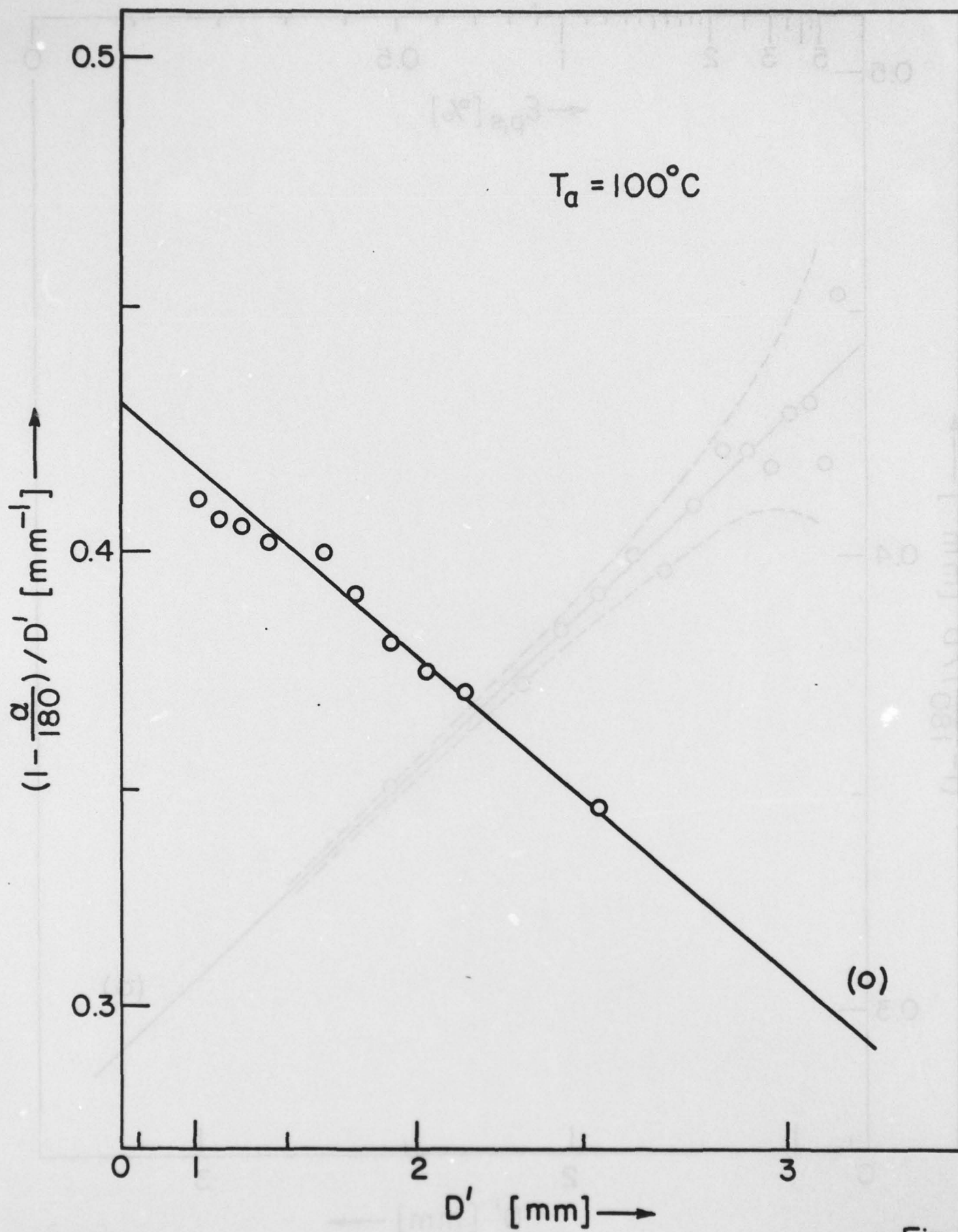


Fig. 3

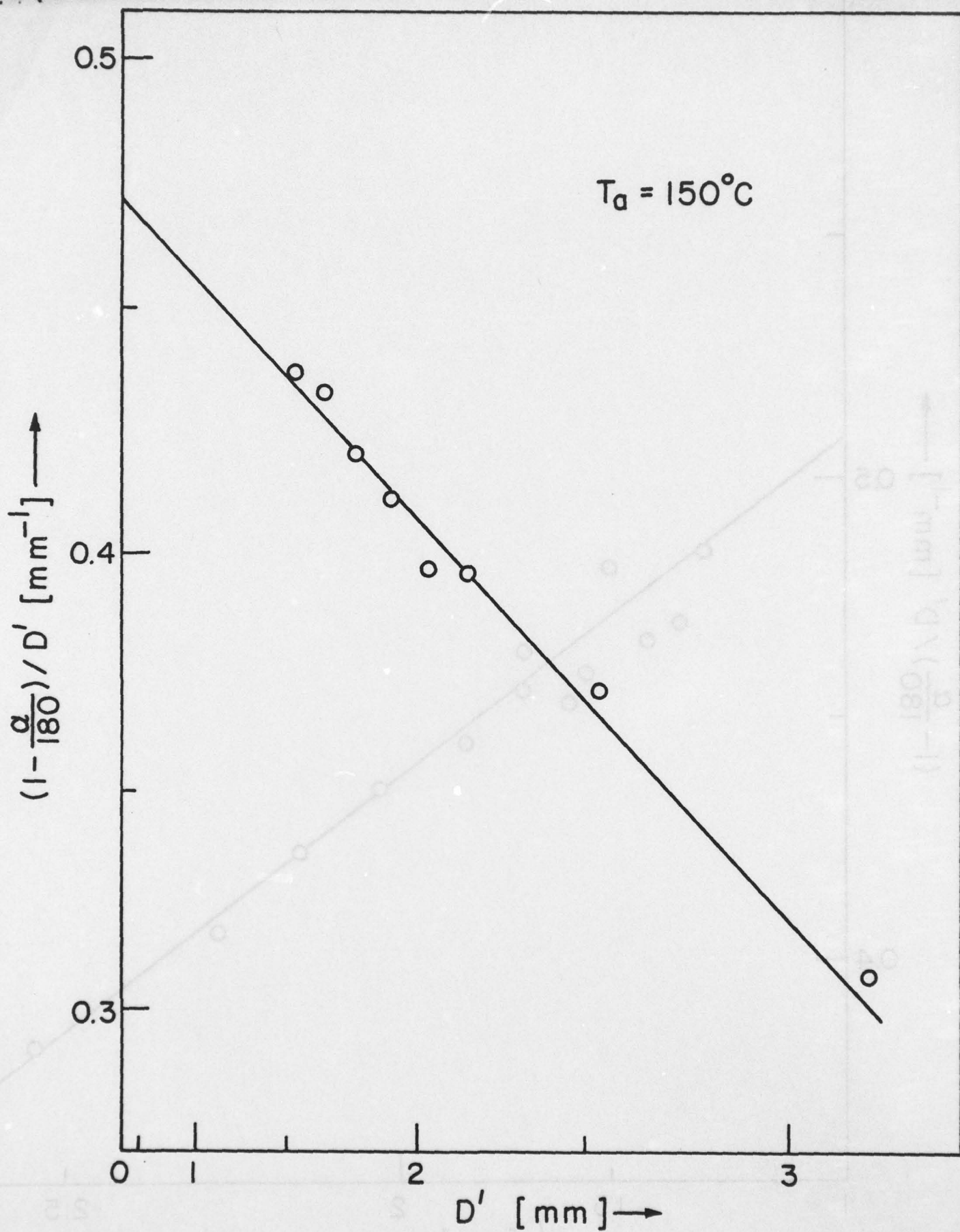


Fig. 4

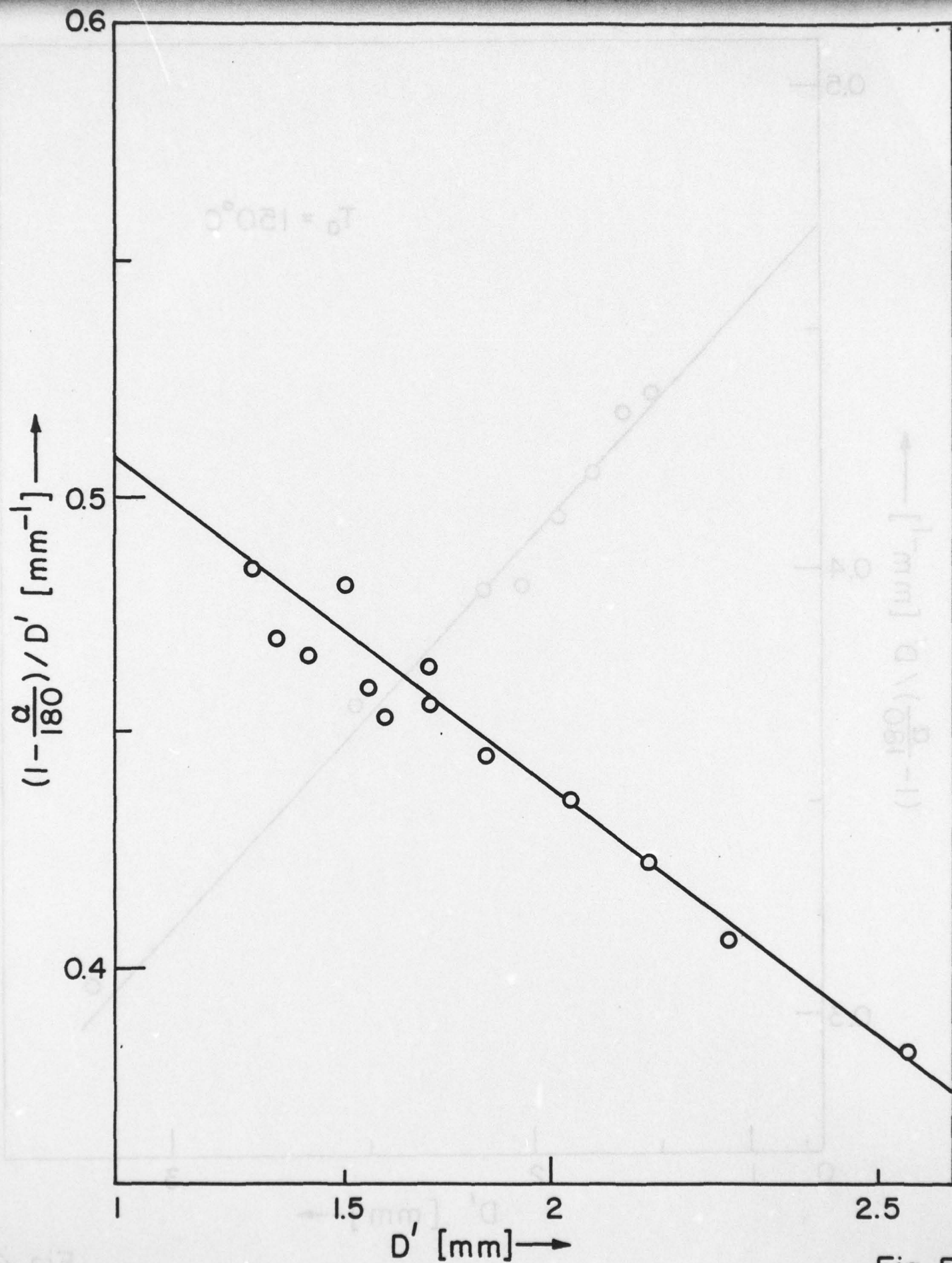


Fig. 5

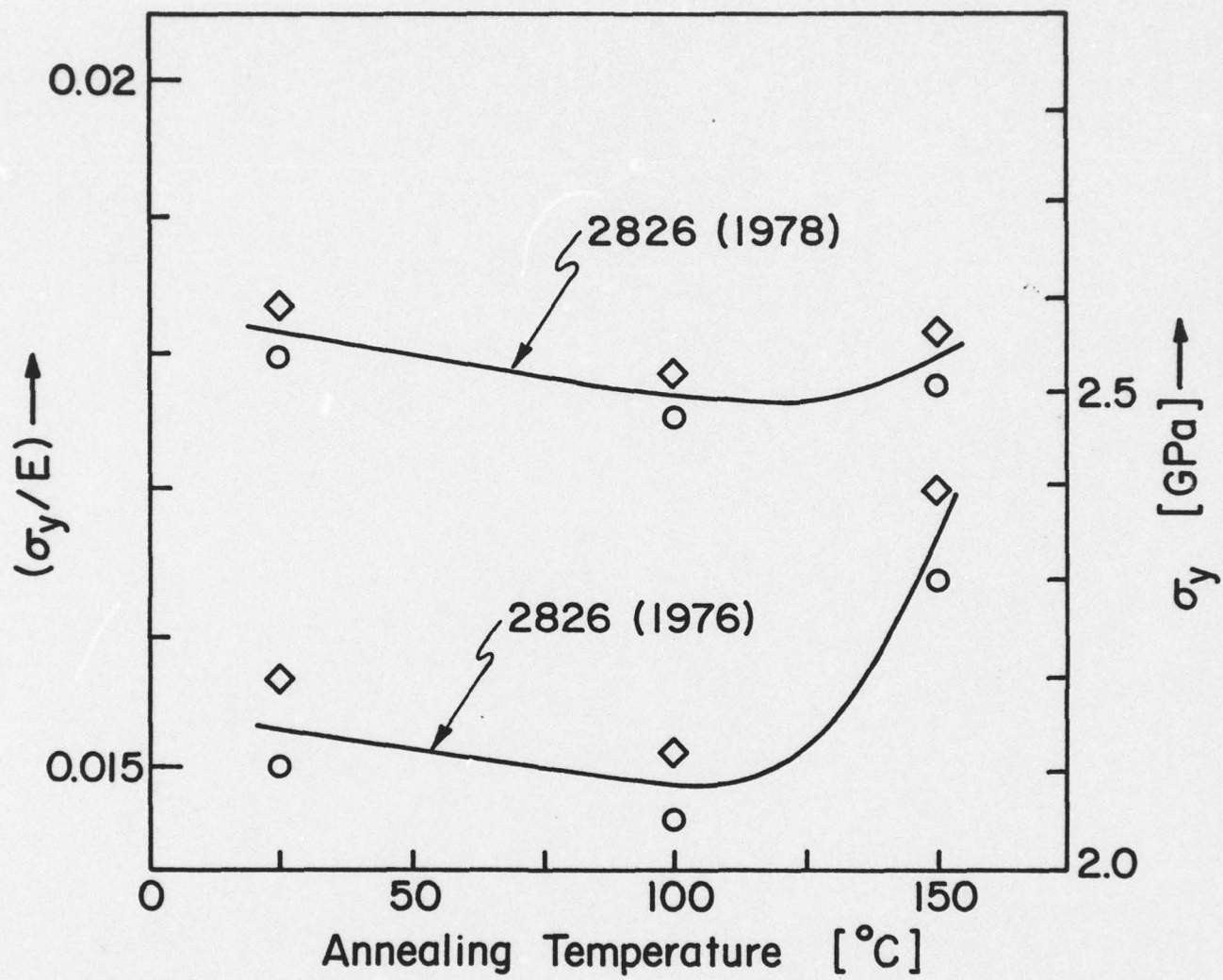


Fig. 6